Size effects in tensile testing of thin cold rolled and annealed Cu foils

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Abstract

Thin copper foils of varying thickness were tested in tension at room temperature. The thickness of the samples was between 10 and 250 μm, length and width were scaled according to the thickness. Two types of samples were tested: as-received samples and annealed samples. The microstructure of the foils was determined in detail by means of several methods. If samples of the same processing condition were compared, there was a clear dependence of the mechanical behavior on the thickness of the foils in the tensile test. When the thickness was reduced from 250 to 10 μm, the fracture strain decreased for the as-received foils from approximately 20% to 0.2% and for the samples with heat treatment from 35% to 15%. It has to be stressed that this size effect occurred in the absence of considerable strain gradients and for samples with a comparable microstructure.

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1. Introduction

In the last decade the interest in the mechanical properties of materials in small dimensions has grown immensely as by means of modern process technologies the fabrication of technical systems at the micro- or even at the nanoscale has become feasible. For a safe and reliable design of such systems it is mandatory to know the mechanical behavior of the materials used. These properties, such as critical stresses and strains, cannot not necessarily be deduced from macroscopic specimens due to two reasons. Firstly, the size of microstructural features, for example the grain size, can be in the order of the size of a structure and, hence, interactions different from that in bulk materials can be expected. Secondly, small structures can be fabricated by means of new processes yielding microstructures not available at the macroscale. These facts generate the need for testing structures in small configurations.

The size dependence of material properties is generally termed “size effect”. In the following, recent findings of size effects in elastic and inelastic properties of engineering materials will be reviewed briefly. The focus of the literature overview will be on the influence of the size on the mechanical properties of materials and not on the type of the test methods.

1.1. Size effects in elastic material behavior

No experimental results could be found which show a size effect on elastic properties in micrometer sized metallic structures. This is confirmed by several studies in which no size dependence of the Young’s modulus was found [1,2]. Conversely, in [3] measurement results are reported which show a decrease of Young’s modulus for Al films thinner than 100 nm.

For more complex materials with large structural features as, for example, foams or bones, there seem to be size effects for elastic properties (see e.g. [4,5]). Generalized continuum theories which incorporate a length scale were developed for explaining these findings such as micropolar and nonlocal elasticity or gradient theories (e.g. in [6–8]).

1.2. Size effects in inelastic material behavior

In inelastic material behavior of crystalline structures other deformation mechanisms than those relevant for elasticity are important, such as dislocation motion and grain boundary sliding [9]. These mechanisms are directly coupled to the microstructure and in particular to the grain size. If the size of a structure is in the
1.2.1. Plastic properties

In [10] thin Cu wires were investigated in tension and torsion (diameters, 12–170 μm). There it was found that the amount of work hardening increased for thinner wires in torsion whereas there was no such effect in tension. This was explained by an increase in the density of geometrically necessary dislocations in torsion, as in torsion there exists a macroscopic strain gradient responsible for the storage of geometrically necessary dislocations in contrast to tension. Similar results were found by [11] for bending of thin Ni foils. Thinner foils required a larger equivalent moment for producing a certain amount of surface strain. Various authors made indentation experiments with different materials such as Al or Cu [12–18]. The general result of these experiments was that the hardness decreased with increasing indentation depth (“indentation size effect”). These findings were explained by hardening due to the strain gradient in the vicinity of the indenter tip. It must be stressed that due to the complexity of the stress field in indentation, size effect studies using this technique are difficult to interpret. An increasing yield stress with decreasing thickness in thin Cu films was found by [19–22] which was explained by models which consider interface dislocations and grain size effects in thin films. An influence of the ratio of thickness to grain size was found for thin foils by [23], the smaller the ratio the smaller the flow stress. Uchic et al. [24] investigated micrometer sized single crystals in compression and found an increase in yield stress with decreasing size. Recent experiments of [25] showed an increase in yield stress for very thin free standing Al, Au and Cu films (thickness 0.2–1 μm) which was interpreted to be independent of the misfit dislocation formation.

1.2.2. Failure properties

Size effects in fracture strength (ultimate tensile strength in tensile testing) were found by a number of authors (e.g. for quasi-brittle materials as concrete and rocks by [26–28], for fcc metals such as Cu by [21,29], for Si by [30]). Although the reason for this behavior differed with the type of material the general finding was that the thinner a structure the higher its ultimate tensile strength. In tensile testing (in a macroscopically uniaxial stress field) a decrease in the strain at fracture with decreasing thickness was found for thin Cu foils by [31–36] and for thin Au and Al foils by [25]. A size dependence of fatigue life was found in tension–tension fatigue of thin Au and Cu wires by [37] as well as of thin Cu foils by [38]. The smaller the size of a structure the higher the number of cycles performed before failure. In [39] multiple step tests of tension–tension fatigue were performed showing that thinner Cu foils had higher plastic strain amplitudes than thicker ones. A change in crack growth rate was observed by [33] in tensile fatigue of thin Al, Cu and Mo foils: foils thinner than 150 μm showed an intermittent crack arrest. A step wise change in crack growth rate was found by [40] who tested small beams fabricated by the LIGA technique (composition NiFe 13% in bending. For completeness, a well known size effect from fracture mechanics is reported. An influence of the sample size on the fracture toughness was found first by [41] who investigated several mm thick Al sheets containing a central crack. In that paper, the sample size dependence of the fracture toughness was attributed to a change in fracture mode (shearing in thin sheets versus flat-tensile fracture mode in thick sheets).

1.3. Size effect theories

Several models were developed for explaining size effects, e.g. strain gradient theories ([42–44], with a reformulation of the latter in [45], and [46]), models with dislocations confined in thin films [47–49], discrete dislocation dynamics theories [50,51], fracture mechanics theories (especially for concrete, [26,52]), and models based on the interactions with microstructural and dimensional constraints [53,54]. Statistical models were proposed already by [55] and later by [56,57].

Different kinds of size effects are reported in the literature. Whereas for standard crystalline materials there are no substantial indications for size effects of the elastic properties, plastic and failure properties seem to be dependent on the size to a certain extent. The general trend is that smaller structures have higher strength and lower ductility than larger ones. However, the variety of size effects reported in literature indicates that various physical phenomena, operating on different length scales, govern mechanical material behavior. Therefore, it is difficult to develop a unified theory which explains all these effects. This is manifested in the variety of models which are proposed to explain size effects. Regarding this complexity, an experimental investigation of the influence of the size of a sample on its mechanical behavior must include a careful determination of the microstructure. In this study, this was tried to be achieved by means of modern material characterization methods. The length scale investigated comprised of Cu foils with thicknesses from 10 to 250 μm while typical grain dimensions varied from a few to about 100 μm.

2. Experiments

Thin rolled copper foils (99.97% purity) of varying thickness (10, 20, 50, 100 and 250 μm) were tested in tension. In the following the experimental key points are discussed, a more detailed description on the experiments performed can be found in [58].

2.1. Samples

2.1.1. Geometry

The examination of size effects requires well defined loading conditions. Therefore, the standard dog bone shape was chosen as the shape for tensile test samples. To guarantee geometrically similar flow conditions, the geometry of the samples, in particular the gauge section, was scaled according to the thickness, H. As a standard, the width was chosen to be W = 20H, the length L = 200H and the radius of the transition arc R = 100H. Fig. 1
2.1.2. Fabrication
Rolled copper foils of 99.97% purity and of 10, 20, 50, 100 and 250 µm thickness served as a base material for the tensile test samples. As the exact rolling procedure was not disclosed to the authors by the manufacturer detailed information on this process cannot be reported. Whereas foils thicker than 50 µm were structured by conventional wet etching techniques, thinner foils had to be fixed to a substrate for stability reasons before processing. Thin Cu foils of 50 mm × 50 mm size were glued on 3 in. Si wafers. For this step, particular care had to be taken to ensure that the foils were perfectly flat. This could be achieved by pressing them overnight with another wafer and a weight. Once the foils were glued on the wafer, the foils could be patterned by standard photolithography (spin-coating, exposing, and developing) taking care of the orientation of the specimen’s axes with respect to the rolling direction of the Cu foil. These axes coincided for the samples tested in this study. The etching of the Cu foils was performed in a 50 °C sodium persulfate (Na₂S₂O₈) solution for several seconds to a few minutes depending on the thickness of the foil. Finally, the foil was removed from the silicon wafer by putting it into a solvent.

With this procedure bars of 50 µm width could be produced. The under etching was negligible in comparison with the width which is 20 times the thickness in the standard design.

2.1.3. Heat treatment
After etching, the samples were ready to be used in a tensile test; these samples were termed “as-received”. In addition, some samples were heat treated in a vacuum at 300 °C for 2 h to study the influence of a change in microstructure (“heat treated” or “annealed” samples).

2.2. Characterization of microstructure and geometry
The microstructure of the samples was analyzed in detail by means of several methods. X-ray diffraction was used for the study of the macrotexture (Seifert XRD3000 PTS Goniometer). Electron backscatter diffraction (EBSD [59]) was applied for a detailed characterization of the microstructure, i.e. the size, the distribution and the orientation of the grains (CAMSCAN CS44LB, OIM 3.5). By means of ion beam slope cutting (IBSC [60]) different sections of the copper foils could be accessed which allowed for a three-dimensional characterization of the microstructure (details in [61]). For confirmation of the EBSD results metallographic samples were made and analyzed by standard light microscopy.

Width and thickness were tested for each sample. The width was measured with a non-contact laser profilometer (Autofokus profilometer 2010, UBM Messtechnik) and the thickness with a contact height gauge (Cary Compare).

3. Results
In the following, a summary of the results of the microstructural investigations and the mechanical behavior under loading is given. Further details can be found in [58].

3.1. Geometry of samples
Thickness and width of each tensile test sample were determined prior to the tensile test. Within one sample these values varied in average less than 1.5%. Variation in both thickness and width was in average less than 6% between different samples.

Table 1 lists the surface roughness parameters of undeformed foils tested parallel to the rolling direction. The surface roughness was in the same order of magnitude for all samples tested meaning that the relative surface roughness was higher for thinner foils. The 20 µm foils had the smallest surface roughness. The applied heat treatment did not have a measurable influence on the surface roughness.
which indicated a strong texture (cube texture (1 0 0)[0 0 1]). Grains which were highly elongated in rolling direction could be observed. Due to gradual changes in

Fig. 3. Result of an EBSD analysis of a 20 μm thick sample. The grain shapes were no longer elongated but were rather equi-axed. As for the as-received samples, the microstructure after annealing was similar for foils of different thickness. In average, the grains had a cubic grain shape and an extension of approximately 15 μm in all directions. Thus, the very thin foils often consisted of only one or two grains per thickness. Annealing twins could be observed frequently. Only a few grains were left in cube orientation. Rolling texture components with (1 1 1) parallel to the rolling direction formed the preferred orientations. However, the texture index was rather low. The difference in texture before and after annealing for the 20 μm thick samples is shown in the pole figures of Fig. 4. Whereas the as-received samples showed high maxima indicating a strong texture, this was not the case any more for the annealed samples.

The results found by EBSD were in good agreement with the metallographic analysis and the X-ray measurements. More detailed information on the microstructure of the as-received and annealed Cu foils can be found in [61].

Tensile tests

Tensile tests were performed using the setups described in Section 2.3. Typically five samples were tested for each thickness and processing condition (more for thinner samples). Fig. 5 shows an engineering stress strain diagram for samples of different thickness and processing conditions. The very thin as-received foils (10 and 20 μm thick) displayed hardly any plastic deformation whereas the 100 μm thick foils showed a ductile behavior typical for copper. The heat treatment induced a strong softening of the material, i.e. the ultimate tensile strength was reduced significantly and the ductile part increased strongly, in particular for the 10 and 20 μm foils. The tensile test results showed good reproducibility: the variation of the stress values between different tests was in the order of 5%. The strain values, in particular the fracture strain, showed higher variation which
Fig. 4. Contoured pole figures for as-received and annealed 20 μm thick unloaded foils (equal contour level values for both). The as-received samples showed a strong cube texture which was hardly present in the annealed samples. The major texture component of the heat treated samples was a weak rolling texture with ⟨111⟩ parallel to the rolling direction.

Fig. 6 summarizes the size dependence of the mechanical behavior of the thin Cu foils tested. The fracture strain, $\varepsilon_{fr}$ and the ultimate tensile strength, $R_m$ were chosen as values for comparison for both the as-received and the annealed samples. A clear trend could be observed regardless of the scattering of the measurement values. Thinner foils had a smaller fracture strain for both the as-received and the annealed samples. When the thickness was reduced from 250 to 10 μm, the fracture strain decreased for the as-received foils from approximately 20% to 0.2% and for the samples with heat treatment from 35% to 15%. The situation was not so clear for the ultimate tensile strength. The annealed samples did not show a strong size dependency at all. The as-received samples, however, showed the tendency that thinner samples were stronger with the 20 μm foil being an exception. It has to be stressed at this point that the microstructure of all samples tested was comparable in the as-received state.

For the same Cu foils as presented in this study, the influence of other parameters such as width and length of a sample, strain rate, surface oxidation and orientation with respect to the rolling direction were examined in [58]. In comparison with the thickness, they only had a minor influence on the mechanical behavior of the foils tested.

Fig. 5. Engineering stress–strain diagram for copper foils of the thicknesses 10, 20 and 100 μm and different processing conditions (sample curves; “annealed” samples were heat treated for 2 h at 300 °C in a vacuum). There was a size dependence of the mechanical behavior. Thinner foils had smaller fracture strains. The annealing induced strong softening, in particular for the 10 and 20 μm foils.
Fig. 6. Size dependence of fracture strain, $A$ (left) and ultimate tensile strength, $R_m$ (right) of thin copper foils tested in tension. Thinner foils had a smaller fracture strain. For the ultimate tensile strength there was no clear trend for the annealed samples. The 10 μm thick foils displayed by far the strongest variation between different samples (S.D. for $A$ 50% and for $R_m$ 10%), for the other foils it was significantly lower (typically for $A$ 15% and for $R_m$ 5%).

Fig. 7. Top view of surfaces of an as-received and annealed 20 μm thick foil after loading. The samples were pulled in the vertical direction (parallel to rolling direction). Deformation structures which were induced by the rolling process prior to the tensile test (e.g. grooves from rolling) were clearly visible. Slip lines at an angle of 45° to the tensile axis could be observed indicating dislocation activity. The as-received foils had a relatively even surface after testing in comparison to the annealed samples.

3.4. Analysis of post-mortem samples

After loading, slip lines could be observed in the sheet plane surfaces for both the as-received and annealed samples (Fig. 7) indicating dislocation motion. In addition, deformation structures, which originated in the rolling process (e.g. grooves from rolling) and which were present before the tensile test, were detectable. The major geometrical modifications took place in the thickness direction what could be observed in SEM-images of fractured specimens. The samples failed by forming a neck in thickness direction with a knife edge type of fracture surface. This type of failure was observed for all samples irrespective of thickness and processing condition. No substantial reduction in thickness could be observed outside the necking region for the 10 and 20 μm as-received foils in contrast to all other samples.

After the tensile test, an increase in surface roughness in the gauge section could be observed for both the as-received and the annealed samples as shown in Fig. 7. The surface of the as-received samples was relatively even outside the necking zone containing some slip lines, whereas the surface of the annealed samples contained slip lines and was relatively uneven not only in the necking zone but also in the entire gauge section (“orange peel effect”). Fig. 8 shows roughness scans for both as-received and annealed 20 μm thick samples after loading which were taken parallel to the tensile axis. For both samples an increase in surface roughness could be observed in the gauge section as there the stresses were maximal. The increase in surface roughness was moderate for the as-received samples (typical roughness parameters increased by 10–20%) and very pronounced for the annealed samples (typical roughness parameters doubled).

4. Discussion

When the tensile test results are analyzed in detail several aspects attract attention: firstly, the strong microstructural changes induced by the heat treatment and the consequential softening of the mechanical response; secondly, the nearly brittle behavior of the 10 and 20 μm as-received samples; thirdly, the size dependence of the mechanical behavior and fourthly, the increase in surface roughness after tensile testing. These facets will be discussed in the following paragraphs.
4.1. Softening induced by heat treatment

As can be seen by a comparison of Figs. 2 and 3, the applied heat treatment at 300 °C for 2 h causes a recrystallization of the grain structure. Thus, there are changes in grain size and shape, texture and dislocation distribution. These changes in microstructure yield a completely different mechanical behavior of the annealed samples. In contrast to the as-received samples, the annealed samples show a stress–strain behavior with a relatively low yield stress and a large amount of work hardening followed by failure caused by necking. The large plastic deformations accompanied by work hardening cause microstructural changes. This can be observed by EBSD analyses of loaded samples [58]. Annealed samples show a moderate elongation of the grains after tensile testing and a strong increase in surface roughness. Furthermore, a thickness reduction in loaded samples can be noticed in the analysis of the fracture surfaces. Obviously, these samples have the ability to work harden substantially as the dislocation density is much lower than in the as-received samples.

Moreover, as the grain size is considerably larger, the fraction of grains at the surface is higher in the annealed samples. These grains can deform more easily due to the reduced constraints at the surface [23]. This is another reason for the increased ductility of the annealed samples.

Thus, comparing as-received and annealed samples shows that, at the length scale investigated (thickness, 10–250 μm), the microstructure has a dominant effect on the mechanical behavior. Besides the well known effects of microstructural parameters such as grain size or dislocation density the importance of nano-twins was recently shown for Cu in [64,65]. There it was found that in samples with nano-twins the ultimate tensile strength was in the order of 1000 MPa whereas the material still showed substantial ductility. However, as these strength values are far larger than the ones found in this study, nano-twins are not considered to play an important role for the deformation behavior of the Cu foils tested.

4.2. Low fracture strain of very thin as-received foils

If the stress–strain curves are analyzed for both, the 10 and 20 μm thick as-received samples, only a very small amount of work hardening can be noticed (see Fig. 5). From a macroscopical point of view, this behavior could be described as being nearly brittle as there hardly seems to be any plastic deformation. Microscopically however, the failure is by no means brittle as there is considerable plastic yielding in the necking zone (Fig. 7). At moderate distances away from the crack tip no microstructural changes are detectable by means of EBSD. This indicates that the plastic deformation field is very inhomogeneous along the gauge section with the plastic strain being localized and maximal at the neck and very small outside of this region. The strain measured in the tensile test is the average strain of the entire gauge section, however. Thus, as the longitudinal extension of the neck is only in the order of the thickness, the large deformations present in this region are averaged out along the total length which is 200 times the thickness. This fact explains the low overall fracture strain, A, which lies in the order of only 0.5%. As can be shown by a simple geometrical model [58], the strain caused by necking is the main contribution to the plastic strain measured in the thin as-received samples. A homogeneous thickness reduction of the gauge section cannot be detected in these samples. The reason for the immediate formation of a neck can be seen in a relatively high dislocation density present in cold rolled copper foils [66]. At very high dislocation densities the work hardening rate is limited by the lower fraction of mobile dislocations [67] and, thus, the material can hardly compensate for geometrical softening as caused by the formation of a neck.

4.3. Size dependence of material behavior

If the left graph of Fig. 6 is analyzed, a clear size dependence can be observed for samples with the same processing
condition (as-received or annealed, respectively): thinner foils display a smaller fracture strain than thicker ones. For the tensile strength, however, no such unambiguous size effect as for the fracture strain can be found (right graph of Fig. 6). For the annealed samples, there does not seem to be a size dependence at all, whereas for the as-received samples the trend is that thinner samples are stronger with the 20 m thick foils being an exception.

For the explanation of these experimental findings, the combination of several mechanisms and effects has to be considered: differences in geometry and microstructure between samples of different thickness and effects of the thickness on the mechanical behavior. The different rolling processes, necessary for the fabrication of foils of varying thickness, are responsible for differences in surface roughness (Table 1) and in microstructure. However, as already reported in Section 3.2, samples of one processing condition turned out to have relatively similar microstructures irrespective of thickness. Variations in microstructure could hardly be educed from the experimental data due to the complicated grain structure present in the as-received foils. Nevertheless, some variance in microstructure for foils of different thickness has to be expected, in particular for the arrangement of dislocation barriers (e.g. grain boundaries and as a consequence grain size) and for the dislocation density.

With respect to the yield stress, the influence of these differences can be roughly estimated by standard models, such as the Hall-Petch relation for the grain size (yield stress inversely proportional to the square root of the grain size) and the Taylor model for the dislocation density (yield stress proportional to the square root of the dislocation density) [67]: a reduction of the grain size by 30% increases the yield stress by 20%, an increase of the dislocation density by 100% increases the yield stress by 40%. Due to the experimental difficulties in the determination of microstructural parameters for the as-received foils, the differences in these can be expected in the order of magnitude mentioned. The calculated differences in yield stress correspond to the variation of the tensile strength found for as-received foils of varying thickness (right graph of Fig. 6). As no pronounced size dependence can be identified for the annealed, recrystallized samples, slight variations in the microstructure of the as-received foils seem to be (at least partially) responsible for the size dependence found for the tensile strength. However, there are other effects which influence the yield stress and hence the tensile strength. In [23,68] it is reported that the flow stress of thin foils is dependent on the number of grains per thickness if there are <10 grains per thickness. This is explained by the relaxed deformation constraint of grains at the surface. Thus, for samples with a constant grain size but varying thickness this results in a size dependent material behavior. In addition, for very thin films (in the order of a few microns and below) there seem to be additional hardening mechanisms depending on the behavior of dislocations in films with an oxide layer [48] and on the number of dislocation sources, active slip systems and dislocation motion paths per thickness [25].

Strain gradient effects are not believed to cause substantial hardening for the samples tested in this study as a tensile load is applied and, hence, the deformation is homogeneous until the formation of a neck (no substantial storage of geometrically necessary dislocations).

The size dependence of the fracture strain can only be partially attributed to the presumed differences in the microstructure of foils with varying thickness. It is much more pronounced as the size dependence observed for the tensile strength and it is also present in the annealed samples. Due to the complexity of the grain structure it is believed that the decrease of the fracture strain with decreasing thickness is a consequence of the combination of various factors. Firstly, as already stated in Section 4.1, the reason for the very low ductility of the thin as-received foils is seen in the limited hardening rate due to a high dislocation density. However, this does not explain the differences in fracture strain for the annealed samples of varying thickness. Secondly, as listed in Table 1 the absolute surface roughness is in the same order of magnitude for all foils tested in the unloaded state. This means that for the samples tested in this study the relative surface roughness is higher for thinner foils. A flaw of a given size, originating e.g. from the rolling process, is a stronger weakening for thinner foils. As a consequence, thinner samples break at smaller macroscopic stresses and smaller strains, either through stress concentration at the flaw or through a reduction of the effective cross-section of the tensile test sample. This finding is supported by the fact, that the variance in the fracture strain is higher for thinner samples which shows the statistical influence of the surface roughness. Thirdly, the smaller the number of grains per cross-section the lower is the probability that a gliding system is oriented properly such that it can be activated. Hence, for samples with similar grain size but different thickness, the ability to undergo plastic deformations is limited in thinner samples and, therefore, smaller fracture strains can be expected.

Thus, there are several reasons for a decrease in the fracture strain with decreasing thickness. However, the importance of each mechanism is difficult to state and has to be investigated separately by varying the grain size and the surface properties. Similar results, i.e. a decrease of the fracture strain with decreasing thickness, were also found by [31,32,34]. However, there are some important differences between these studies and the current one, in particular, from an experimental point of view. In this work, geometrically similar samples were tested whereas in the other studies the samples only differed in thickness (the importance of testing geometrically similar specimens is discussed in [58]). Moreover, no information on the surface roughness is given in the other studies. Thus, the importance of this factor cannot be assessed.

To sum up, it is believed that the combination of various effects is responsible for the size dependence of the fracture strain in tensile testing determined in this study: slight differences in microstructure of samples, influence of surface roughness and number of activated gliding systems. The following effects are not seen as important for the interpretation of the size effects found in this study: strain gradients (homogeneous deformation in tensile testing), stress state (assumed to be plane stress, details in [58]) and grain boundary sliding (no grain boundary sliding could be observed in as-received samples, only minor grain boundary sliding in annealed samples).
4.4. Surface roughening through loading

The surface roughening in the gauge section after tensile testing (Fig. 8) can be explained by two factors. Firstly, many slip bands can be observed on the surface of the deformed as-received and annealed samples. This indicates that plastic deformation is accompanied by dislocation slip in a substantial amount. These slip bands increase microscopically the surface roughness. For the annealed samples there is a second factor. As there are only a few grains per cross-section these grains are not strongly constrained in their deformation as they are bounded by free surfaces. As shown in Figs. 3 and 4, these grains are, in general, oriented differently. Thus, in order to accommodate the plastic deformations these grains can rotate out of their initial position by gliding along the grain boundaries (uneven plastic flow). This mechanism yields the surface roughening which can be observed on a larger scale than the roughening due to slip bands.

5. Conclusions

Thin rolled copper foils of varying thickness with a scaled geometry and comparable microstructure were tested in tension. Some samples were annealed for 2 h at 300 °C in a vacuum prior to the tensile tests. The most important result of the tensile tests is that the thickness of the foils has an influence on the mechanical behavior in the size regime studied. When the thickness is reduced from 250 to 10 μm, the fracture strain decreases for the as-received foils from approximately 20% to 0.2% and for the samples with heat treatment from 35% to 15%. In terms of stress-strain behavior there was no clear trend for the annealed samples, the as-received samples showed the tendency that thinner samples were stronger. The annealing induced a strong softening meaning a strong increase in ductility for the very thin foils (e.g. for the 10 μm thick foils the fracture strain increased from 0.2% to 15%). The size dependence of the fracture strain can be qualitatively explained by the combination of several factors. Firstly, the work hardening rate is limited in thinner samples due to a high dislocation density. Secondly, thinner samples are more sensitive to geometrical imperfections stemming from the surface roughness. Thirdly, for a constant grain size there are fewer grains in thinner samples which could result in a smaller number of activated gliding systems.

It has to be stressed that the size dependence found in this work resulted from a tensile test, i.e. a test where no considerable strain gradients occur. Experimental verification of size effects in loading situations, where no strain gradients are present, is scarce (e.g. [25,39]). This work also shows that, for the explanation of the effects observed, a thorough examination of the microstructure of the samples tested is mandatory. As the influence of many parameters has to be taken into account in detail, the experimental study of size effects turns out to be a complicated topic. There are several mechanisms which influence the mechanical behavior and these mechanisms can even have opposite effects.

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References


